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SPECIFICATION

HIGH-STIFFNESS HIGH-STRENGTH THIN STEEL SHEET AND METHOD FOR PRODUCING THE SAME

RELATED APPLICATIONS

[0001] This is a §371 of International Application No. PCT/JP2005/06327, with an international filing date of March 31, 2005 (WO 2005/095664 A1, published October 13, 2005), which is based on Japanese Patent Application Nos. 2004-107040, filed March 31, 2004, and 2004-346620, filed November 30, 2004.

TECHNICAL FIELD

[0002] This disclosure relates to a high-stiffness high-strength thin steel sheet suitable mainly as a vehicle body for automobiles and a method for producing the same. Moreover, the high-stiffness high-strength thin steel sheet is a column-shaped structural member having a thickness susceptibility index of the stiffness near to 1 such as center pillar, locker, side flame, cross member or the like of the automobile and is widely suitable for applications requiring a stiffness.

BACKGROUND

[0003] As a result of recent heightened interest in global environment problems, the exhaust emission control is conducted even in the automobiles, and hence the weight reduction of the vehicle body in the automobile is a very important matter. For this end, it is effective to attain the weight reduction of the vehicle body by increasing the strength of the steel sheet to reduce the thickness thereof.

[0004] Recently, the increase of the strength in the steel sheet is considerably advanced, and hence the use of thin steel sheets having a thickness of less than 2.0 mm is increasing. To further reduce the weight by the increase of the strength, it is indispensable to simultaneously control the deterioration of the stiffness in parts through the thinning of the thickness. Such a problem of deteriorating the stiffness of the parts through the thinning

of the thickness in the steel sheet is actualized in steel sheets having a tensile strength of not less than 590 MPa, and particularly this problem is serious in steel sheets having a tensile strength of not less than 700 MPa.

[0005] In general, to increase the stiffness of the parts, it is effective to change the shape of the parts, or to increase the number of welding points or change the welding condition such as changeover to laser welding or the like in the spot-welded parts. However, when these parts are used in the automobile, there are problems that it is not easy to change the shape of the parts in a limited space inside the automobile, and the change of the welding conditions causes the increase of the cost and the like.

[0006] Consequently, to increase the stiffness of the parts without changing the shape of the parts or the welding conditions, it becomes effective to increase the Young's modulus of the material used in the parts.

[0007] In general, the stiffness of the parts under the same shape of parts and welding conditions is represented by a product of Young's modulus of the material and geometrical moment of inertia of the part. Further, the geometrical moment of inertia can be expressed so as to be approximately proportionate to t^{λ} when the thickness of the material is t. In this case, λ is a thickness susceptibility index and is a value of 1-3 in accordance with the shape of the parts. For example, in case of one plate shape such as panel parts for the automobile, λ is a value near to 3, while in case of column-shape such as structural parts, λ is a value near to 1.

[0008] When λ of the parts is 3, if the thickness is made small by 10% while equivalently maintaining the stiffness of the parts, it is required to increase the Young's modulus of the material by 37%, while when λ of the parts is 1, if the thickness is made small by 10%, it may be enough to increase the Young's modulus by 11%.

[0009] That is, in case of the parts having λ near to 1 such as column-shaped parts, it is very effective to increase the Young's modulus of the steel sheet itself for the weight reduction. Particularly, in case of steel sheets having a

high strength and a small thickness, it is strongly demanded to highly increase the Young's modulus of the steel sheet.

[0010] In general, the Young's modulus is largely dependent upon the texture and is known to become high in a closest direction of atom. Therefore, it is effective to develop {112}<110> to develop an orientation advantageous for the Young's modulus of steel being a body-centered cubic lattice in a steel making process comprising the rolling through rolls and the heat treatment, whereby the Young's modulus can be increased in a direction perpendicular to the rolling direction.

[0011] There have hitherto been variously examined steel sheets by controlling the texture to increase the Young's modulus.

[0012] For example, JP-A-H05-255804 discloses a technique wherein a steel obtained by adding Nb or Ti to an extremely low carbon steel is hot-rolled at a rolling reduction at Ar_3-(Ar_3+150°C) of not less than 85% to promote transformation from non-crystallized austenite to ferrite to thereby render the texture of ferrite at the stage of the hot-rolled sheet into {311}<011> and {332}<113>, which is an initial orientation and is subjected to a cold rolling and a recrystallization annealing to render {211}<011> into a main orientation to thereby increase the Young's modulus in a direction perpendicular to the rolling direction.

[0013] Also, JP-A-H08-311541 discloses a method for producing a hot rolled steel sheet having an increased Young's modulus in which Nb, Mo and B are added to a low carbon steel having a C content of 0.02-0.15% and the rolling reduction at Ar_3-950°C is made to not less than 50% to develop [211]<011>.

[0014] Further, JP-A-H09-53118 discloses a method for producing a hot rolled steel sheet in which Si and Al are added to a low carbon steel having a C content of not more than 0.05% to enhance Ar_3 transformation point and the rolling reduction below Ar_3 transformation point in the hot rolling is made to not less than 60% to increase Young's modulus in a direction perpendicular to the rolling direction.

[0015] However, the aforementioned techniques have the following problems.

[0016] In the technique disclosed in JP-A-H05-255804, the Young's modulus of the steel sheet is increased by using the extremely low carbon steel having a C content of not more than 0.01% to control the texture, but the tensile strength is low as about 450 MPa at most, so that there is a problem in the increase of the strength by applying this technique.

[0017] In the technique disclosed in JP-A-H08-311541, since the C content is as high as 0.02-0.15%, it is possible to increase the strength, but as the target steel sheet is the hot rolled steel sheet, the control of the texture through cold working can not be utilized, and hence there are problems that it is difficult to further increase the Young's modulus but also it is difficult to stably produce high-strength steel sheets having a thickness of less than 2.0 mm through low-temperature finish rolling.

[0018] Further, in the technique disclosed in JP-A-H09-53118, the crystal grains are coarsened by conducting the rolling at the ferrite zone, so that there is a problem that the workability is considerably deteriorated.

[0019] Thus, the increase of the Young's modulus in the steel sheet by the conventional techniques is targeted to hot rolled steel sheets having a thick thickness or soft steel sheets, so that it is difficult to increase the Young's modulus of high-strength thin steel sheet having a thickness of not more than 2.0 mm by using the above conventional techniques.

[0020] As a strengthening mechanism for increasing the tensile strength of the steel sheet to not less than 590 MPa, there are mainly a precipitation strengthening mechanism and a transformation texture strengthening mechanism.

[0021] When the precipitation strengthening mechanism is used as the strengthening mechanism, it is possible to increase the strength while suppressing the lowering of the Young's modulus of the steel sheet as far as possible, but the following difficulty is accompanied. That is, when utilizing the precipitation strengthening mechanism for finely precipitating, for example,

a carbonitride of Ti, Nb or the like, in the hot rolled steel sheet, the increase of the strength is attained by conducting the fine precipitation in the coiling after the hot rolling, but in the cold rolled steel sheet, the coarsening of the precipitate can not be avoided at the step of recrystallization annealing after the cold rolling and it is difficult to increase the strength through the precipitation strengthening.

[0022] When utilizing the transformation texture strengthening mechanism as the strengthening mechanism, there is a problem that the Young's modulus of the steel sheet lowers due to strain included in a low-temperature transformation phase such as bainite phase, martensite phase or the like.

[0023] It could, therefore, be helpful to provide a high-stiffness high-strength thin steel sheet having a tensile strength of not less than 590 MPa, preferably not less than 700 MPa, a Young's modulus of not less than 230 GPa, preferably not less than 240 GPa and a thickness of not more than 2.0 mm as well as an advantageous method for producing the same.

SUMMARY

[0024] We provide:

(I) A high-stiffness high-strength thin steel sheet comprising C: 0.02-0.15%, Si: not more than 1.5%, Mn: 1.0-3.5%, P: not more than 0.05%, S: not more than 0.01%, Al: not more than 1.5%, N: not more than 0.01% and Ti: 0.02-0.50% as mass%, provided that C, N, S and Ti contents satisfy the relationships of the following equations (1) and (2):

$$Ti^* = Ti-(47.9/14)\times N-(47.9/32.1)\times S \geq 0.01 \quad \cdots\cdots (1)$$

$$0.01 \leq C-(12/47.9)\times Ti^* \leq 0.05 \quad \cdots\cdots (2)$$

and the remainder being substantially iron and inevitable impurities, and having a texture comprising a ferrite phase as a main phase and having a martensite phase at an area ratio of not less than 1%, and having a tensile strength of not less than 590 MPa and a Young's modulus of not less than 230 GPa.

[0025] (II) A high-stiffness high-strength thin steel sheet according to the item (I), which further contains one or two of Nb: 0.005-0.04% and V: 0.01-0.20% as mass% in addition to the above composition and satisfies the relationships of the above equation (1) and the following equation (3) instead of the equation (2):

$$0.01 \leq C-(12/47.9)\times Ti^{*}-(12/92.9)\times Nb-(12/50.9)\times V \leq 0.05 \cdots (3).$$

[0026] (III) A high-stiffness high-strength thin steel sheet according to the item (I) or (II), which further contains one or more of Cr: 0.1-1.0%, Ni: 0.1-1.0%, Mo: 0.1-1.0%, Cu: 0.1-2.0% and B: 0.0005-0.0030% as mass% in addition to the above composition.

[0027] (IV) A method for producing a high-stiffness high-strength thin steel sheet comprising subjecting a starting material of steel comprising C: 0.02-0.15%, Si: not more than 1.5%, Mn: 1.0-3.5%, P: not more than 0.05%, S: not more than 0.01%, Al: not more than 1.5%, N: not more than 0.01% and Ti: 0.02-0.50% as mass%, provided that C, N, S and Ti contents satisfy the relationships of the following equations (1) and (2):

$$Ti^{*} = Ti-(47.9/14)\times N-(47.9/32.1)\times S \geq 0.01 \cdots\cdots (1)$$

$$0.01 \leq C-(12/47.9)\times Ti^{*} \leq 0.05 \cdots\cdots (2)$$

to a hot rolling step under conditions that a total rolling reduction below 950°C is not less than 30% and a finish rolling is terminated at 800-900°C, coiling the hot rolled sheet below 650°C, pickling, subjecting to a cold rolling at a rolling reduction of not less than 50%, raising a temperature to 780-900°C at a temperature rising rate from 500°C of 1-30°C/s to conduct soaking, and then cooling at a cooling rate up to 500°C of not less than 5°C/s to conduct annealing.

[0028] (V) A method for producing a high-stiffness high-strength thin steel sheet according to the item (IV), wherein the starting material of steel further contains one or two of Nb: 0.005-0.04% and V: 0.01-0.20% as mass% in addition to the above composition and satisfies the relationships of the above equation (1) and the following equation (3) instead of the equation (2):

$$0.01 \leq C-(12/47.9)\times Ti^{*}-(12/92.9)\times Nb-(12/50.9)\times V \leq 0.05 \quad \cdots \quad (3)$$

[0029] (VI) A method for producing a high-stiffness high-strength thin steel sheet according to the item (IV) or (V), wherein the staring material of steel further contains one or more of Cr: 0.1-1.0%, Ni: 0.1-1.0%, Mo: 0.1-1.0%, Cu: 0.1-2.0% and B: 0.0005-0.0030% as mass% in addition to the above composition.

[0030] It is possible to provide a high-stiffness high-strength thin steel sheet having a tensile strength of not less than 590 MPa and a Young's modulus of not less than 230 GPa.

[0031] That is, the starting material of low carbon steel added with Mn and Ti is roll-reduced below 950°C in the hot rolling to promote the transformation from non-recrystallized austenite to ferrite and then cold rolled to develop a crystal orientation useful for the improvement of Young's modulus and thereafter a low-temperature transformation phase suppressing the lowering of the Young's modulus is produced and a greater amount of ferrite phase useful for the improvement of the Young's modulus is retained in the cooling stage by the control of the heating rate in the annealing step and the soaking at two-phase region, whereby the thin steel sheet satisfying higher strength and higher Young's modulus can be produced, which develops an effective effect in industry.

[0032] Further explaining in detail, the starting material of low carbon steel added with Mn and Ti is roll-reduced at an austenite low-temperature region in the hot rolling to increase the non-recrystallized austenite texture having a crystal orientation of {112}<111>, and subsequently the transformation from the non-recrystallized austenite of {112}<111> to ferrite is promoted in the cooling stage to develop ferrite orientation of {113}<110>.

[0033] In the cold rolling after the coiling and pickling, the rolling is carried out at a rolling reduction of not less than 50% to turn the crystal orientation of {113}<110> to {112}<110> useful for the improvement of the Young's modulus, and in the temperature rising stage at the subsequent annealing step, the

temperature is raised from 500°C to the soaking temperature at a heating rate of 1-30°C/s to promote the recrystallization of ferrite having an orientation of {112}<110> and provide a two-phase region at a state of partly retaining the non-recrystallized grains of {112}<110>, whereby the transformation from the non-recrystallized ferrite of {112}<110> to austenite can be promoted.

[0034] Further, in the transformation from austenite phase to ferrite phase at the cooling after the soaking, ferrite grains having an orientation of {112}<110> is grown to enhance the Young's modulus, while the steel enhancing the hardenability by the addition of Mn is cooled at a rate of not less than 5°C/s to produce the low-temperature transformation phase, whereby it is attempted to increase the strength.

[0035] Moreover, the low-temperature transformation phase is produced by retransforming the austenite phase transformed from ferrite having an orientation of {112}<110> during the cooling, so that {112}<110> can be also developed even in the crystal orientation of the low-temperature transformation phase.

[0036] Thus, the Young's modulus is enhanced by developing {112}<110> of ferrite phase, and particularly {112}<110> is increased in the orientation of the low-temperature transformation phase largely exerting on the lowering of the Young's modulus, whereby the strength can be increased by the formation of the low-temperature transformation phase and the lowering of the Young's modulus accompanied with the formation of the low-temperature transformation phase can be largely suppressed.

BRIEF DESCRIPTION OF THE DRAWINGS

[0037] FIG. 1 is a graph showing an influence of a total rolling reduction below 950°C on Young's modulus;

FIG. 2 is a graph showing an influence of a final temperature in hot finish rolling on Young's modulus;

FIG. 3 is a graph showing an influence of a coiling temperature on Young's modulus;

FIG. 4 is a graph showing an influence of a rolling reduction in cold rolling on Young's modulus; and

FIG. 5 is a graph showing an influence of an average temperature rising rate from 500°C to soaking temperature in annealing on Young's modulus.

DETAILED DESCRIPTION

[0038] The high-stiffness high-strength thin steel sheet is a steel sheet having a tensile strength of not less than 590 MPa, preferably not less than 700 MPa, a Young's modulus of not less than 230 GPa, preferably not less than 240 GPa, and a thickness of not more than 2.0 mm. Moreover, the steel sheet includes steel sheets subjected to a surface treatment such as galvanization inclusive of alloying, zinc electroplating or the like in addition to the cold rolled steel sheet.

[0039] The reason of limiting the chemical composition in the steel sheet will be described below. Moreover, the unit for the content of each element in the chemical composition of the steel sheet is "% by mass", but it is simply shown by "%" unless otherwise specified.

[0040] C: 0.02-0.15%

C is an element stabilizing austenite and can largely contribute to increase the strength by enhancing the hardenability at the cooling stage in the annealing after the cold rolling to largely promote the formation of the low-temperature transformation phase. Further, C can contribute to increase the Young's modulus by promoting the transformation of ferrite grains having {112}<110> after the cold rolling from the non-recrystallized ferrite to austenite in the temperature rising stage at the annealing step.

[0041] To obtain such effects, the C content is required to be not less than 0.02%, preferably not less than 0.05%, more preferably not less than 0.06%. On the other hand, when the C content exceeds 0.15%, the fraction of hard low-temperature transformation phase becomes large, and the strength of the steel is extremely increased but also the workability is deteriorated. Also, the greater amount of C suppresses the recrystallization of the orientation useful for the

increase of the Young's modulus at the annealing step after the cold rolling. Further, the greater amount of C brings about the deterioration of the weldability.

[0042] Therefore, the C content is required to be not more than 0.15%, preferably not more than 0.10%.

[0043] Si: not more than 1.5%

Si raises the Ar_3 transformation point in the hot rolling, so that when the rolling is terminated at 800-900°C, if Si is contained in an amount exceeding 1.5%, the rolling at austenite region becomes difficult and the crystal orientation required for the increase of the Young's modulus can not be obtained. Also, the greater amount of Si deteriorates the weldability of the steel sheet but also promotes the formation of fayalite on a surface of a slab in the heating at the hot rolling step to accelerate the occurrence of surface pattern so-called as a red scale. Furthermore, in case of using as a cold rolled steel sheet, Si oxide produced on the surface deteriorates the chemical conversion processability, while in case of using as a galvanized steel sheet, Si oxide produced on the surface induces non-plating. Therefore, the Si content is required to be not more than 1.5%. Moreover, in case of steel sheets requiring the surface properties or the galvanized steel sheet, the Si content is preferable to be not more than 0.5%.

[0044] Also, Si is an element stabilizing ferrite and promotes the ferrite transformation at the cooling stage after the soaking of two-phase region in the annealing step after the cold rolling to enrich C in austenite, whereby austenite can be stabilized to promote the formation of the low-temperature transformation phase. For this end, the strength of steel can be increased, if necessary. To obtain such an effect, the Si content is desirable to be not less than 0.2%.

[0045] Mn: 1.0-3.5%

Mn is one important element. Mn has an action of suppressing the recrystallization of worked austenite in the hot rolling. Also, Mn can promote

the transformation from the non-recrystallized austenite to ferrite to develop {113}<110> and improve the Young's modulus in the subsequent cold rolling and annealing steps.

[0046] Furthermore, Mn as an austenite stabilizing element lowers Ac_1 transformation point in the temperature rising stage at the annealing step after the cold rolling to promote the transformation from the non-recrystallized ferrite to austenite, and can develop the orientation useful for the improvement of the Young's modulus to control the lowering of the Young's modulus accompanied with the formation of the low-temperature transformation phase with respect to the orientation of the low-temperature transformation phase produced in the cooling stage after the soaking.

[0047] Also, Mn enhances the hardenability in the cooling stage after the soaking and annealing at the annealing step to largely promote the formation of the low-temperature transformation phase, which can largely contribute to the increase of the strength. Further, Mn acts as a solid-solution strengthening element, which can contribute to the increase of the strength in steel. To obtain such an effect, the Mn content is required to be not less than 1.0%, preferably not less than 1.5%.

[0048] On the other hand, when the Mn content exceeds 3.5%, Ac_3 transformation point is excessively lowered in the temperature rising stage at the annealing step after the cold rolling, so that the recrystallization of ferrite phase at the two-phase region is difficult and it is required to raise the temperature up to an austenite single-phase region above Ac_3 transformation point. As a result, ferrite of {112}<110> orientation useful for the increase of the Young's modulus obtained by the recrystallization of worked ferrite can not be developed to bring about the lowering of the Young's modulus. Further, the greater amount of Mn deteriorates the weldability of the steel sheet. Moreover, the greater amount of Mn enhances the deformation resistance of steel in the hot rolling to increase the rolling load, which causes the difficulty in the operation. Therefore, the Mn content is not more than 3.5%.

[0049] P: not more than 0.05%

Since P segregates in the grain boundary, if the P content exceeds 0.05%, the ductility and toughness of the steel sheet lower but also the weldability is deteriorated. In case of using the alloyed galvanized steel sheet, the alloying rate is delayed by P. Therefore, the P content is required to be not more than 0.05%. On the other hand, P is an element effective for the increase of the strength as a solid-solution strengthening element and has an action of promoting the enrichment of C in austenite as a ferrite stabilizing element. In the steel added with Si, it has also an action of suppressing the occurrence of red scale. To obtain these actions, the P content is preferable to be not less than 0.01%.

[0050] S: not more than 0.01%

S considerably lowers the hot ductility to induce hot tearing and considerably deteriorate the surface properties. Further, S hardly contributes to the strength but also forms coarse MnS as an impurity element to lower the ductility and drill-spreading property. These problems become remarkable when the S content exceeds 0.01%, so that it is desirable to reduce the S content as far as possible. Therefore, the S content is not more than 0.01%. From a viewpoint of improving the drill-spreading property, it is preferable to be not more than 0.005%.

[0051] Al: not more than 1.5%

Al is an element useful for deoxidizing steel to improve the cleanness of the steel. However, Al is a ferrite stabilizing element, and largely raises the Ar_3 transformation of the steel, so that when the finish rolling is terminated at 800-900°C, if the Al content exceeds 1.5%, the rolling at austenite region becomes difficult to suppress the development of the crystal orientation required for the increase of the Young's modulus. Therefore, the Al content is required to be not more than 1.5%. From this viewpoint, Al is preferable to be made lower, and further preferable to be limited to not more than 0.1%. On the other hand, Al as a ferrite forming element promotes the formation of ferrite in the cooling stage after the soaking at the two-phase region in the

annealing step after the cold rolling to enrich C in austenite, whereby austenite can be stabilized to promote the formation of the low-temperature transformation phase. As a result, the strength of the steel can be enhanced, if necessary. To obtain such an effect, the Al content is desirable to be not less than 0.2%.

[0052] N: not more than 0.01%

N is a harmful element because slab breakage is accompanied in the hot rolling to cause surface defect. When the N content exceeds 0.01%, the occurrence of slab breakage and surface defect becomes remarkable. Further, when a carbonitride forming element such as Ti, Nb or the like is added, N forms a coarse nitride at a high temperature to suppress the effect by the addition of the carbonitride forming element. Therefore, the N content is required to be not more than 0.01%.

[0053] Ti: 0.02-0.50%

Ti is a most important element. That is, Ti controls the recrystallization of worked austenite at the finish rolling step in the hot rolling to promote the transformation from the non-recrystallized austenite to ferrite and develop {113}<110>, and can increase the Young's modulus at the subsequent cold rolling and annealing steps. Also, the recrystallization of worked ferrite is suppressed in the temperature rising stage at the annealing step after the cold rolling to promote the transformation from the non-recrystallized ferrite to austenite, and the orientation useful for the increase of the Young's modulus can be developed with respect to the orientation of the low-temperature transformation phase produced in the cooling stage after the soaking to suppress the lowering of the Young's modulus accompanied with the formation of the low-temperature transformation phase. Further, a fine carbonitride of Ti can contribute to the increase of the strength. To obtain such actions, the Ti content is required to be not less than 0.02%, preferably not less than 0.03%.

[0054] On the other hand, when the Ti content exceeds 0.50%, all the carbonitride can not be solid-soluted in the re-heating at the usual hot rolling step and a coarse carbonitride remains, and hence the effect of suppressing the recrystallization of worked austenite at the hot rolling step or the effect of suppressing the recrystallization of worked ferrite at the annealing step after the cold rolling can not be obtained. Also, even if the hot rolling of the slab after the continuous casting is started as it is without conducting the re-heating after the continuously cast slab is cooled, when the Ti content exceeds 0.50%, the improvement of the effect of suppressing the recrystallization is not recognized and also the increase of the alloy cost is brought about. Therefore, the Ti content is required to be not more than 0.50%, preferably not more than 0.20%.

[0055] The contents of C, N, S and Ti are required to satisfy the relationship of the following equations (1) and (2):

$$Ti^* = Ti-(47.9/14)\times N-(47.9/32.1)\times S \geq 0.01 \quad \cdots\cdots \quad (1)$$

$$0.01 \leq C-(12/47.9)\times Ti^* \leq 0.05 \quad \cdots\cdots \quad (2).$$

[0056] Ti is liable to easily form coarse nitride and sulfide at a high temperature region. The formation of such nitride and sulfide brings about the reduction of the effect of suppressing the recrystallization through the addition of Ti. Therefore, the amount of $Ti^* = Ti-(47.9/14)\times N-(47.9/32.1)\times S$ as an amount of Ti not fixed as the nitride and sulfide is required to be not less than 0.01%, preferably not less than 0.02%.

[0057] If C not fixed as a carbonitride is existent in an amount exceeding 0.05%, the introduction of strain in the cold rolling becomes non-uniform and further the recrystallization of the orientation useful for the increase of the Young's modulus is suppressed, so that the C amount not fixed as the carbide calculated by $(C-(12/47.9)\times Ti^*)$ is required to be not more than 0.05%. On the other hand, when the C amount not fixed as the carbide is less than 0.01%, the C content in austenite decreases in the annealing at two-phase region after the cold rolling to suppress the formation of martensite phase after the cooling

and hence it is difficult to increase the strength. Therefore, the amount of C-(12/47.9)×Ti* as the C amount not fixed as the carbide is 0.01-0.05%.

[0058] Moreover, the term "the remainder being substantially iron and inevitable impurities" used herein means that steels containing slight amounts of other elements without damaging the steel. In case of further increasing the strength, one or two of Nb: 0.005-0.04% and V: 0.01-0.20% and one or more of Cr, Ni, Mo, Cu and B may be added, if necessary, in addition to the above definition of the chemical composition

[0059] Nb: 0.005-0.04%

Nb is an element contributing to the increase of the strength by forming a fine carbonitride. Also, it is an element contributing to the increase of the Young's modulus by suppressing the recrystallization of worked austenite at the finish rolling step in the hot rolling to promote the transformation from the non-recrystallized austenite to ferrite. To obtain such actions, the Nb content is preferable to be not less than 0.005%. On the other hand, when the Nb content exceeds 0.04%, the rolling load considerably increases in the hot rolling and cold rolling and the difficulty is accompanied in the production, so that the Nb content is preferably not more than 0.04%, more preferably not more than 0.01%.

[0060] V: 0.01-0.20%

V is an element contributing to the increase of the strength by forming a fine carbonitride. Since it has such an action, the V content is preferable to be not less than 0.01%. On the other hand, when the V content exceeds 0.20%, the effect of increasing the strength by the amount exceeding 0.20% is small and the increase of the alloy cost is caused. Therefore, the V content is preferable to be 0.01-0.20%.

[0061] When Nb and/or V are included in addition to Ti, the contents of C, N, S, Ti, Nb and V are required to satisfy the relationship of the following equation (3) instead of the equation (2):

$$0.01 \leq C-(12/47.9)\times Ti^{*}-(12/92.9)\times Nb-(12/50.9)\times V \leq 0.05 \cdots (3).$$

[0062] Nb and V form the carbide to decrease the C content not fixed as the carbide. Therefore, to render the C content not fixed as the carbide into 0.01-0.05%, when Nb and/or V are added, the value of C-(12/47.9)×Ti*-(12/92.9)×Nb-(12/50.9)×V is required to be 0.01-0.05%.

[0063] Cr: 0.1-1.0%

Cr is an element enhancing the hardenability by suppressing the formation of cementite and can largely contribute to the increase of the strength by largely promoting the formation of the low-temperature transformation phase in the cooling stage after the soaking at the annealing step. Further, the recrystallization of worked austenite is suppressed in the hot rolling step to promote the transformation from non-recrystallized austenite to ferrite and develop {113}<110>, and the Young's modulus can be increased at the subsequent cold rolling and annealing steps. To obtain such an effect, Cr is preferable to be included in an amount of not less than 0.1%. On the other hand, when the Cr content exceeds 1.0%, the above effect is saturated and the alloy cost increases, so that Cr is preferable to be included in an amount of not more than 1.0%. Moreover, when the thin steel sheet is used as a galvanized steel sheet, the oxide of Cr produced on the surface induces the non-plating, so that Cr is preferable to be included in an amount of not more than 0.5%.

[0064] Ni: 0.1-1.0%

Ni is an element stabilizing austenite to enhance the hardenability, and can largely contribute to the increase of the strength by largely promoting the formation of the low-temperature transformation phase in the cooling stage after the soaking at the annealing step. Further, Ni as an austenite stabilizing element lowers Ac_1 transformation point in the temperature rising stage at the annealing step after the cold rolling to promote the transformation from the non-recrystallized ferrite to austenite, and develops the orientation useful for the increase of the Young's modulus with respect to the orientation of the low-temperature transformation phase produced in the cooling stage after the

soaking, whereby the lowering of the Young's modulus accompanied with the formation of the low-temperature transformation phase can be suppressed. Moreover, Ni suppresses the recrystallization of worked austenite in the hot rolling to promote the transformation from the non-recrystallized austenite to ferrite to thereby develop {113}<110>, whereby the Young's modulus can be increased at the subsequent cold rolling and annealing steps. In case of the steel added with Cu, the surface defect is induced by cracking accompanied with the lowering of the hot ductility in the hot rolling, but the occurrence of the surface defect can be controlled by composite addition of Ni. To obtain such an action, Ni is preferable to be included in an amount of not less than 0.1%.

[0065] On the other hand, when the Ni content exceeds 1.0%, Ac_3 transformation point is extremely lowered in the temperature rising stage at the annealing step after the cold rolling and the recrystallization of ferrite phase at the two-phase region is difficult, and hence it is required to raise the temperature up to austenite single phase region above Ac_3 transformation point. As a result, ferrite of orientation obtained by the recrystallization of worked ferrite and useful for the increase of the Young's modulus can not be developed to bring about the decrease of the Young's modulus. And also, the alloy cost increases. Therefore, Ni is preferable to be included in an amount of not more than 1.0%.

[0066] Mo: 0.1-1.0%

Mo is an element enhancing the hardenability by making small the mobility of the interface, and can largely contribute to the increase of the strength by largely promoting the formation of the low-temperature transformation phase in the cooling stage at the annealing step after the cold rolling. Further, the recrystallization of worked austenite can be suppressed, and the transformation from the non-recrystallized austenite to ferrite is promoted to develop {113}<110> and the Young's modulus can be increased at the subsequent cold rolling and annealing steps. To obtain such an action, Mo

is preferable to be included in an amount of not less than 0.1%. On the other hand, when the Mo content exceeds 1.0%, the above effect is saturated and the alloy cost increases, so that Mo is preferable to be included in an amount of not more than 1.0%.

[0067] B: 0.0005-0.0030%

B is an element suppressing the transformation from austenite phase to ferrite phase to enhance the hardenability, and can largely contribute to the increase of the strength by largely promoting the formation of the low-temperature transformation phase in the cooling stage at the annealing step after the cold rolling. Further, the recrystallization of worked austenite can be suppressed, and the transformation from the non-recrystallized austenite to ferrite is promoted to develop {113}<110> and the Young's modulus can be increased at the subsequent cold rolling and annealing steps. To obtain such an effect, B is preferable to be included in an amount of not less than 0.0005%. On the other hand, when the B content exceeds 0.0030%, the deformation resistance in the hot rolling is enhanced to increase the rolling load and the difficulty is accompanied in the production, so that B is preferable to be included in an amount of not more than 0.0030%.

[0068] Cu: 0.1-2.0%

Cu is an element enhancing the hardenability, and can largely contribute to the increase of the strength by largely promoting the formation of the low-temperature transformation phase in the cooling stage at the annealing step after the cold rolling. To obtain such an effect, Cu is preferable to be included in an amount of not less than 0.1%. On the other hand, when the Cu content exceeds 2.0%, the hot ductility is lowered and the surface defect accompanied with the cracking in the hot rolling is induced and the hardening effect by Cu is saturated, so that Cu is preferable to be included in an amount of not more than 2.0%.

[0069] The reason on the limitation of the texture will be described below.

[0070] In the thin steel sheet, it is required to have a texture comprising a ferrite phase as a main phase and having a martensite phase at an area ratio of not less than 1%.

[0071] The term "ferrite phase as a main phase" used herein means that the area ratio of the ferrite phase is not less than 50%.

[0072] Since the ferrite phase is less in the strain, useful for the increase of the Young's modulus, excellent in the ductility and good in the workability, the texture is required to be the ferrite phase as a main phase.

[0073] Also, to render the tensile strength of the steel sheet into not less than 590 MPa, it is required that the low-temperature transformation phase as a hard phase is formed in a portion other than the ferrite phase as a main phase or a so-called second phase to provide a composite phase. At this moment, the feature that a hard martensite phase among the low-temperature transformation phases is particularly existent in the texture is advantageous because the fraction of the second phase for obtaining the target tensile strength level is made small and the fraction of ferrite phase is made large, whereby the increase of the Young's modulus is attained and further the workability can be improved. For this end, the martensite phase is required to be not less than 1% as an area ratio to the whole of the texture. To obtain the strength of lot less than 700 MPa, the area ratio of the martensite phase is preferable to be not less than 16%.

[0074] The texture of the steel sheet is preferable to be a texture comprising ferrite phase and martensite phase, but there is no problem that phases other than the ferrite phase and martensite phase such as bainite phase, residual austenite phase, pearlite phase, cementite phase and the like are existent at the area ratio of not more than 10%, preferably not more than 5%. That is, the sum of area ratios of ferrite phase and martensite phase is preferably not less than 90%, more preferably not less than 95%.

[0075] Next, the reason on the production conditions limited for obtaining the high-stiffness high-strength thin steel sheet and preferable production conditions will be explained.

[0076] The composition of the starting material of steel used in the production method is the same as the composition of the aforementioned steel sheet, so that the description of the reason on the limitation of the starting material of steel is omitted.

[0077] The thin steel sheet can be produced by successively conducting a hot rolling step of subjecting the starting material of steel having the same composition as the composition of the steel sheet to a hot rolling to obtain a hot rolled sheet, a cold rolling step of subjecting the hot rolled sheet after pickling to a cold rolling to obtain a cold rolled sheet, and an annealing step of attaining the recrystallization and composite texture in the cold rolled sheet.

[0078] (Hot rolling step)

Finish rolling: total rolling reduction below 950°C is not less than 30%, and the rolling is terminated at 800-900°C.

[0079] In the final rolling at the hot rolling step, the rolling is conducted at a lower temperature to develop a non-recrystallized austenite texture having a crystal orientation of $\{112\}<111>$, and the $\{112\}<111>$ non-recrystallized austenite can be transformed to ferrite in the subsequent cooling stage to develop ferrite orientation of $\{113\}<110>$. This orientation advantageously acts to the improvement of the Young's modulus in the formation of the texture at the subsequent cold rolling and annealing steps. To obtain such an action, it is required that the total rolling reduction below 950°C (total rolling reduction) is not less than 30% and further the finish rolling is terminated below 900°C. On the other hand, when the final temperature of the finish rolling is lower than 800°C, the rolling load considerably increases due to the increase of the deformation resistance and the difficulty is accompanied in the production. Therefore, the final temperature of the finish rolling is required to be not lower than 800°C.

[0080] Coiling temperature: not higher than 650°C

When the coiling temperature after the finish rolling exceeds 650°C, the carbonitride of Ti is coarsened and the effect of suppressing the recrystallization of ferrite becomes small in the temperature rising stage at the annealing step after the cold rolling and it is difficult to transform the non-recrystallized ferrite into austenite. As a result, the orientation of the low-temperature transformation phase transformed in the cooling stage after the soaking can not be controlled, and the Young's modulus is largely lowered by the low-temperature transformation phase having such a strain. Therefore, the coiling temperature after the finish rolling is required to be not higher than 650°C. Moreover, when the coiling temperature is too low, a great amount of the hard low-temperature transformation phase is produced and the load in the subsequent cold rolling is increased to cause the difficulty in the production, so that it is preferable to be not lower than 400°C.

[0081] (Cold rolling step)

Cold rolling is carried out at a rolling reduction of not less than 50% after the pickling.

[0082] After the hot rolling step, the pickling is carried out for removing scale formed on the surface of the steel sheet. The pickling may be conducted according to the usual manner. Thereafter, the cold rolling is conducted. By the cold rolling at a rolling reduction of not less than 50% can be turned the orientation of {113}<110> developed on the hot rolled steel sheet to an orientation of {112}<110> effective for the increase of the Young's modulus. Thus, as the orientation of {112}<110> is developed by the cold rolling, the orientation of {112}<110> in ferrite is enhanced in the texture after the subsequent annealing step and further the orientation of {112}<110> is developed in the low-temperature transformation phase, whereby the Young's modulus can be increased. To obtain such an effect, the rolling reduction in the cold rolling is required to be not less than 50%.

[0083] (Annealing step)

Temperature rising rate from 500°C to soaking temperature: 1-30°C/s, Soaking temperature: 780-900°C

[0084] The temperature rising rate at the annealing step is an important process condition. In the course of raising the temperature to a soaking temperature of two-phase region or a soaking temperature of 780-900°C at the annealing step, the recrystallization of ferrite having an orientation of {112}<110> is promoted, while a part of ferrite grains having an orientation of {112}<110> is arrived to a two-phase region at a non-recrystallized state, whereby the transformation from the non-recrystallized ferrite having an orientation of {112}<110> can be promoted. Therefore, the Young's modulus can be increased by promoting the growth of ferrite grains having an orientation of {112}<110> when austenite is transformed into ferrite in the cooling after the soaking. Further, when the strength is increased by producing the low-temperature transformation phase, austenite phase transformed from ferrite having an orientation of {112}<110> is re-transformed in the cooling, so that {112}<110> can be also developed with respect to the crystal orientation of the low-temperature transformation phase. By developing {112}<110> of ferrite phase is increased the Young's modulus, while {112}<110> is particularly developed in the orientation of the low-temperature transformation phase largely influencing the lowering of the Young's modulus, whereby the lowering of the Young's modulus accompanied with the formation of the low-temperature transformation phase can be suppressed while forming the low-temperature transformation phase. When austenite is transformed from the non-recrystallized ferrite while promoting the recrystallization of ferrite in the temperature rising stage, an average temperature rising rate largely exerting on the recrystallization behavior from 500°C to 780-900°C as a soaking temperature is required to be 1-30°C/s. Also, the reason why the soaking temperature is 780-900°C is due to the fact that when it is lower than 780°C, the non-recrystallized texture remains, while when it exceeds 900°C, the amount of austenite produced becomes large and it

is difficult to develop ferrite having an orientation of {112}<110> useful for the increase of the Young's modulus.

[0085] Moreover, the soaking time is not particularly limited, but it is preferable to be not less than 30 seconds for forming austenite, while it is preferable to be not more than about 300 seconds because the production efficiency is deteriorated as the time is too long.

[0086] Cooling rate to 500°C after soaking: not less than 5°C/s

In the cooling stage after the soaking, it is required to form the low-temperature transformation phase containing martensite for increasing the strength. Therefore, an average cooling rate to 500°C after the soaking is required to be not less than 5°C/s.

[0087] Steel having a chemical composition in accordance with the target strength level is first melted. As the melting method can be properly applied a usual converter process, an electric furnace process and the like. The molten steel is cast into a slab, which is subjected to a hot rolling as it is or after the cooling and heating. After the finish rolling under the aforementioned finish conditions in the hot rolling, the steel sheet is coiled at the aforementioned coiling temperature and then subjected to usual pickling and cold rolling. As to the annealing, the temperature is raised under the aforementioned condition, and in the cooling after the soaking, the cooling rate can be increased within a range of obtaining a target low-temperature transformation phase. Thereafter, the cold rolled steel sheet may be subjected to an overaging treatment, or may be passed through a hot dip zinc in case of producing as a galvanized steel sheet, or further in case of producing as an alloyed galvanized steel sheet, a re-heating may be conducted up to a temperature above 500°C for the alloying treatment.

EXAMPLES

[0088] The following examples are given in illustration and are not intended as limitations thereof.

[0089] At first, a steel A having a chemical composition shown in Table 1 is melted in a vacuum melting furnace of a laboratory and cooled to room temperature to prepare a steel ingot (steel raw material).

[0090] Table 1

Kind of steel	Chemical composition										Remarks
	C	Si	Mn	P	S	Al	N	Ti	Ti*	SC	
A	0.06	0.2	2.5	0.02	0.001	0.03	0.002	0.12	0.11	0.03	Acceptable example

(Note) $Ti^* = Ti-(47.9/14)\times N-(47.9/32.1)\times S$

$SC = C-(12/47.9)\times Ti^*$

[0091] Thereafter, the hot rolling, pickling, cold rolling and annealing are successively conducted in the laboratory. The basic production conditions are as follows. After the steel ingot is heated at 1250°C for 1 hour, the hot rolling is conducted under conditions that the total rolling reduction below 950°C is 40% and the final rolling temperature (corresponding to a final temperature of finish rolling) is 860°C to obtain a hot rolled sheet having a thickness of 4.0 mm. Thereafter, the coiling condition (corresponding to a coiling temperature of 600°C) is simulated by leaving the hot rolled sheet up to 600°C and keeping in a furnace of 600°C for 1 hour and then cooling in the furnace. The thus obtained hot rolled sheet is pickled and cold-rolled at a rolling reduction of 60% to a thickness of 1.6 mm. Then, the temperature of the cold rolled sheet is raised at 10°C/s on average up to 500°C and further from 500°C to a soaking temperature of 820°C at 5°C/s on average. Next, the soaking is carried out at 820°C for 180 seconds, and thereafter the cooling is carried out at an average cooling rate of 10°C/s up to 500°C, and further the temperature of 500°C is kept for 80 seconds, and then the sheet is cooled in air.

[0092] In this experiment, the following conditions are further individually changed under the above production conditions as a basic condition. That is, the experiment is carried out under the basic condition except for the individual changed conditions that the total rolling reduction below 950°C is 20-60% and the final temperature of the hot finish rolling is 800-920°C and the

coiling temperature is 500-670°C and the rolling reduction of the cold rolling is 40-75% and the average temperature rising rate from 500°C to the soaking temperature (820°C) in the annealing is 0.5-35°C/s.

[0093] From the sample after the annealing is cut out a test specimen of 10 mm x 120 mm in a direction perpendicular to the rolling direction as a longitudinal direction, which is finished to a thickness of 0.8 mm by a mechanical polishing and a chemical polishing for removing strain, and thereafter a resonance frequency of the sample is measured by using a lateral vibration type internal friction measuring device to calculate a Young's modulus therefrom. With respect to the sheet subjected to a temper rolling of 0.5%, a tensile test specimen of JIS No. 5 is cut out in the direction perpendicular to the rolling direction and subjected to a tensile test. Further, the sectional texture is observed by a scanning type electron microscope (SEM) after the corrosion with Nital to judge the kind of the texture, while three photographs are shot at a visual region of 30 μm x 30 μm and then area ratios of ferrite phase and martensite phase are measured by an image processing to determine an average value of each phase as an area ratio (fraction) of each phase.

[0094] As a result, the values of the mechanical characteristics under the basic condition in the experiment according to the production method are Young's modulus E: 242 GPa, TS: 780 MPa, El: 23%, fraction of ferrite phase: 67% and fraction of martensite phase: 28%, from which it is clear that the thin steel sheet has an excellent balance of strength-ductility and a high Young's modulus.

[0095] Moreover, the remainder of the texture other than ferrite phase and martensite phase is either of bainite phase, residual austenite phase, pearlite phase and cementite phase.

[0096] Then, the relationship between the production conditions and Young's modulus is explained based on the above test results with reference to the drawings. Even in any experimental conditions, the tensile strength is 730-

820 MPa, and the fraction of ferrite phase is 55-80%, the fraction of martensite phase is 17-38%, and the remainder of the texture is either of bainite phase, residual austenite phase, pearlite phase and cementite phase.

[0097] In FIG. 1 is shown influences of the total rolling reduction below 950°C upon Young's modulus, respectively. When the total rolling reduction is not less than 30% being the acceptable range, the Young's modulus indicates an excellent value of not less than 230 GPa.

[0098] In FIG. 2 is shown an influence of the final temperature of the hot finish rolling upon the Young's modulus. When the final temperature is not higher than 900°C being the acceptable range, the Young's modulus indicates an excellent value of not less than 230 GPa.

[0099] In FIG. 3 is shown an influence of the coiling temperature upon the Young's modulus. When the coiling temperature is not higher than 650°C being the acceptable range, the Young's modulus indicates an excellent value of not less than 230 GPa.

[0100] In FIG. 4 is shown an influence of the rolling reduction of the cold rolling upon the Young's modulus. When the rolling reduction is not less than 50% being the acceptable range, the Young's modulus indicates an excellent value of not less than 230 GPa.

[0101] In FIG. 5 is shown an influence of the average temperature rising rate from 500°C to the soaking temperature of 820°C in the annealing upon the Young's modulus. When the temperature rising rate is 1-30°C/s being the acceptable range, the Young's modulus indicates an excellent value of not less than 230 GPa.

[0102] Furthermore, steels B-Z and AA-AI having a chemical composition as shown in Table 2 are melted in a vacuum melting furnace of a laboratory and cooled to room temperature to prepare a steel ingot (steel raw material). Thereafter, it is successively subjected to the hot rolling, pickling, cold rolling and annealing under conditions shown in Table 3, respectively. After the steel ingot is heated at 1250°C for 1 hour, the hot rolling is carried out at various

rolling temperatures to obtain a hot rolled sheet having a thickness of 4.0 mm. Then, the coiling condition after a target coiling temperature is simulated by keeping in a furnace of the coiling temperature for 1 hour and then cooling in the furnace. The hot rolled sheet is pickled, cold-rolled at various rolling reductions to a thickness of 0.8-1.6 mm, and the temperature is raised up to 500°C at 10°C on average and further up to a target soaking temperature at an average temperature rising rate shown in Table 3. After the soaking is carried out at the soaking temperature for 180 seconds, the cooling is carried out at various average cooling rates shown in Table 3, and the sheet is kept at 500°C for 80 seconds and then cooled to room temperature in air.

[0103] In Table 4 are shown characteristics obtained by the aforementioned tests. At this moment, the residual texture other than ferrite phase and martensite phase in the tables is either of bainite phase, residual austenite phase, pearlite phase and cementite phase.

[0104] Table 2

Kind of Steel	Chemical composition (mass%)									Ti*	SC	Remarks
	C	Si	Mn	P	S	Al	N	Ti	other components			
B	0.02	0.20	2.5	0.02	0.001	0.02	0.003	0.06	–	0.05	0.01	Acceptable Steel
C	0.04	0.01	3.0	0.01	0.002	0.03	0.002	0.03	–	0.02	0.03	Acceptable Steel
D	0.01	0.20	2.5	0.02	0.001	0.03	0.003	0.05	–	0.04	0.00	Comparative Steel
E	0.11	0.20	2.5	0.02	0.002	0.04	0.004	0.15	–	0.13	0.08	Comparative Steel
F	0.07	0.20	2.5	0.02	0.002	0.04	0.004	0.05	–	0.03	0.06	Comparative Steel
G	0.06	0.50	2.0	0.03	0.001	0.05	0.005	0.10	–	0.08	0.04	Acceptable Steel
H	0.06	1.50	3.5	0.03	0.001	0.05	0.005	0.15	–	0.13	0.03	Acceptable Steel
I	0.06	0.20	1.5	0.03	0.001	0.05	0.005	0.15	–	0.13	0.03	Acceptable Steel
J	0.06	0.20	1.4	0.03	0.001	0.05	0.002	0.15	–	0.14	0.02	Acceptable Steel
K	0.06	0.30	3.6	0.03	0.001	0.05	0.001	0.12	–	0.12	0.03	Comparative Steel
L	0.05	0.20	2.5	0.03	0.002	0.10	0.002	0.10	–	0.09	0.03	Acceptable Steel
M	0.04	0.30	2.0	0.01	0.003	0.50	0.003	0.08	–	0.07	0.02	Acceptable Steel
N	0.04	0.50	3.0	0.01	0.001	1.50	0.001	0.09	–	0.09	0.02	Acceptable Steel
O	0.05	0.10	2.5	0.01	0.001	0.03	0.006	0.03	–	0.01	0.05	Acceptable Steel
P	0.05	0.10	2.5	0.01	0.001	0.03	0.001	0.02	–	0.02	0.05	Acceptable Steel
Q	0.06	0.20	2.5	0.02	0.001	0.03	0.002	0.05	Nb:0.03	0.04	0.05	Acceptable Steel
R	0.06	0.20	2.5	0.02	0.001	0.03	0.002	0.05	Nb:0.03, V:0.10	0.04	0.02	Acceptable Steel
S	0.04	0.01	3.0	0.02	0.002	0.02	0.002	0.08	Cr:0.3	0.07	0.02	Acceptable Steel
T	0.08	0.03	2.0	0.01	0.002	0.03	0.003	0.15	Ni:0.2	0.14	0.05	Acceptable Steel
U	0.05	0.20	1.5	0.02	0.001	0.03	0.003	0.12	Mo:0.2	0.11	0.02	Acceptable Steel
V	0.04	0.30	2.8	0.03	0.001	0.02	0.003	0.08	Cu:0.3	0.07	0.02	Acceptable Steel
W	0.06	0.20	2.5	0.02	0.001	0.03	0.002	0.13	B:0.0010	0.12	0.03	Acceptable Steel
X	0.05	0.20	2.5	0.01	0.001	0.02	0.003	0.10	Nb:0.03, Mo:0.15	0.09	0.02	Acceptable Steel
Y	0.06	0.30	2.4	0.01	0.002	0.02	0.001	0.05	Cr:0.2, Ni:0.2	0.04	0.05	Acceptable Steel
Z	0.05	0.20	2.4	0.02	0.001	0.02	0.002	0.04	Nb:0.04, Mo:0.15, B:0.0010	0.03	0.04	Acceptable Steel
AA	0.07	0.20	2.9	0.02	0.001	0.02	0.003	0.05	Nb:0.02, V:0.05, Cr:0.1, Ni:0.02, Mo:0.2, Cu:0.3, B:0.0015	0.04	0.05	Acceptable Steel
AB	0.11	0.01	1.7	0.01	0.001	0.02	0.001	0.23	–	0.23	0.05	Acceptable Steel
AC	0.14	0.20	1.8	0.01	0.001	0.03	0.001	0.40	–	0.40	0.04	Acceptable Steel
AD	0.16	0.01	2.0	0.02	0.001	0.03	0.001	0.07	–	0.07	0.14	Comparative Steel
AE	0.05	0.02	1.1	0.01	0.001	0.03	0.001	0.08	–	0.08	0.03	Acceptable Steel
AF	0.05	0.03	0.9	0.01	0.001	0.02	0.001	0.05	–	0.05	0.04	Comparative Steel
AG	0.06	0.03	2.0	0.01	0.001	0.04	0.001	0.03	Nb:0.01	0.03	0.05	Acceptable Steel
AH	0.07	0.20	1.8	0.01	0.001	0.03	0.001	0.07	Nb:0.01	0.07	0.05	Acceptable Steel
AI	0.05	0.01	2.0	0.02	0.001	0.03	0.002	0.01	–	0.00	0.05	Comparative Steel

(Note) $Ti^* = Ti - (47.9/14) \times N - (47.9/32.1) \times S$

$SC = C - (12/47.9) \times Ti^* - (12/92.9) \times Nb - (12/50.9) \times V$

[0105] Table 3

Kind of steel	Hot rolling conditions: Total rolling reduction below 950°C (%)	Hot rolling conditions: Final temperature of finish rolling (°C)	Hot rolling conditions: Coiling temperature (°C)	Cold rolling condition: Rolling reduction (%)	Annealing conditions: Temperature rising rate from 500°C (°C/s)	Annealing conditions: Soaking temperature (°C)	Annealing conditions: Cooling rate up to 500°C (°C/s)	Remarks
B	50	830	550	65	10	840	15	Example
C	45	840	500	70	15	800	20	Example
D	50	850	530	70	8	800	25	Comparative Example
E	45	870	600	60	10	810	20	Comparative Example
F	50	850	550	65	10	820	10	Comparative Example
G	35	880	650	70	10	800	20	Example
H	45	860	540	75	10	860	15	Example
I	50	830	550	70	10	870	30	Example
J	50	830	550	70	10	870	30	Example
K	55	800	500	60	12	810	15	Comparative Example
L	40	870	550	70	10	870	20	Example
M	45	880	540	75	30	870	25	Example
N	45	890	550	70	15	880	20	Example
O	50	830	550	65	10	820	15	Example
P	50	820	500	75	10	830	10	Example
Q	40	850	550	60	10	820	15	Example
R	40	850	550	60	10	820	15	Example
S	50	840	570	80	25	840	30	Example
T	30	870	600	60	10	835	12	Example
U	40	850	580	65	15	840	10	Example
V	45	845	550	65	15	820	17	Example
W	35	860	600	60	15	830	10	Example
X	30	840	550	65	10	860	15	Example
Y	40	850	570	65	15	840	10	Example
Z	40	860	600	60	10	840	15	Example
AA	30	870	630	60	10	845	13	Example
AB	35	850	650	60	10	830	10	Example
AC	45	870	630	55	10	840	10	Example
AD	40	860	600	60	10	860	15	Comparative Example
AE	40	840	600	65	15	860	10	Example
AF	30	820	600	60	10	850	10	Comparative Example
AG	35	840	550	50	10	810	10	Example
AH	35	860	580	50	10	830	15	Example
AI	40	870	600	60	10	840	10	Comparative Example

[0106] Table 4

Kind of steel	Steel texture		Mechanical Properties			Remarks
	Fraction of martensite phase (%)	Fraction of ferrite phase (%)	TS (MPa)	El (%)	E (GPa)	
B	5	93	610	30	251	Example
C	11	87	680	25	245	Example
D	0	100	540	33	252	Comparative Example
E	70	30	1200	11	218	Comparative Example
F	45	55	1030	15	222	Comparative Example
G	35	61	830	20	243	Example
H	25	73	850	18	245	Example
I	20	80	760	24	243	Example
J	15	85	740	25	235	Example
K	40	60	850	18	225	Comparative Example
L	25	70	760	22	243	Example
M	22	72	700	20	245	Example
N	20	75	750	21	245	Example
O	30	65	800	21	234	Example
P	35	60	810	20	233	Example
Q	35	65	820	20	245	Example
R	25	72	780	22	247	Example
S	25	75	760	23	245	Example
T	35	65	890	18	242	Example
U	26	71	790	23	243	Example
V	19	81	750	25	245	Example
W	30	70	900	17	243	Example
X	30	68	890	17	248	Example
Y	40	60	920	16	243	Example
Z	35	65	980	15	245	Example
AA	45	55	1030	14	243	Example
AB	30	65	920	16	235	Example
AC	25	70	940	15	231	Example
AD	70	25	1100	11	210	Comparative Example
AE	5	90	630	30	230	Example
AF	2	96	570	33	223	Comparative Example
AG	20	80	780	19	233	Example
AH	25	70	780	20	238	Example
AI	20	75	750	20	208	Comparative Example

[0107] In the steel D, the C content is as small as 0.01%, and the fraction of martensite is 0%, and TS is smaller than the acceptable range. In the steel E, the C content not fixed as the carbide (SC) is as high as 0.08% and the fraction of ferrite phase is as small as 30%, and the Young's modulus is smaller than the acceptable range. In the steel F, SC is as high as 0.06%, and the Young's modulus is smaller than the acceptable range. In the steel K, the Mn content is as high as 3.6%, and the Young's modulus is smaller than the acceptable range. In the steel AD, the C content is as high as 0.16% and SC is as high as 0.14% and the fraction of ferrite phase is as small as 25%, and the Young's modulus is smaller than the acceptable range. In the steel AF, the Mn content is as low as 0.9%, and TS and the Young's modulus are smaller than the acceptable range. In the steel AI, the Ti content is as low as 0.01% and Ti* is as small as 0.00%, and the Young's modulus is smaller than the acceptable range.

[0108] With respect to the other steels, all items are within the acceptable range, and TS and Young's modulus satisfy the acceptable range.

INDUSTRIAL APPLICABILITY

[0109] It is possible to provide high-stiffness high-strength thin steel sheets having a tensile strength of not less than 590 MPa and a Young's modulus of not less than 230 GPa.